



Microstructure and room-temperature mechanical property of TC21 alloy treated by continuous multistep hydrogenation at different temperatures

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Abstract: Microstructure and room-temperature mechanical property of TC21 titanium alloy treated by continuous multistep hydrogenation at different temperatures were studied by OM, XRD, TEM and compressive test. Results showed that the size of α phase decreased obviously and β phase became the main phase when the hydrogenation temperature exceeded 1073 K. The content of δ hydride increased first and then decreased gradually with the increase of hydrogenation temperature. The strain softening phenomenon appeared during the compression process of TC21 titanium alloy when the hydrogenation temperature exceeded 1073 K and became obvious with the increase of hydrogenation temperature. The ultimate compression of TC21 titanium alloy decreased first and then increased with the increase of hydrogenation temperature, and increased by 67.51% compared with that of the as-received TC21 titanium alloy when the hydrogenation temperature was 1123 K.

Key words: TC21 alloy; continuous multistep hydrogenation treatment; room-temperature mechanical property; pressure–composition isotherm

1 Introduction

In recent years, light metals have played an increasingly important role in many engineering fields [1–4]. As one of the most promising light metals, titanium alloy is widely used in the fields of aerospace, chemical engineering and biomedicine because of its low density, high strength, good biocompatibility, excellent erosion resistance, and fatigue property [5–8]. TC21 alloy is a kind of $\alpha+\beta$ titanium alloy with high strength, high toughness and damage tolerance property [9–11], and has been widely used in aerospace as an important engineering material [12–14]. However, similar to

most titanium alloys, TC21 alloy has low plasticity and high deformation resistance at room temperature, which makes it difficult to process at room temperature [15–17]. Luckily, some researchers found that hydrogen could change the phase composition and improve the mechanical properties of titanium alloys [18–20]. GUO et al [21] studied the effect of hydrogen treatment on microstructures evolution and mechanical properties of titanium alloy fabricated by selective laser melting, and found that the room-temperature plasticity of hydrogenated Ti–6Al–4V samples after solution treatment increased with the increase of hydrogen content (in the range of 0.45–0.80 wt.%). YUAN et al [22] studied the influence of hydrogen

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content on the room-temperature compressive properties of Ti–6Al–4V alloy at high strain rates, and found that hydrogen had favorable effects on the compressive properties of Ti–6Al–4V alloy, and the plasticity increased by 47.0% when 0.2 wt.% hydrogen was introduced. However, most hydrogenation processes adopted to improve the room-temperature plasticity of titanium alloys included solid solution and quenching [23–25], leading to the inconvenience for large-size specimens. Therefore, we proposed a hydrogenation process, continuous multistep hydrogenation treatment (CMHT), to simplify the hydrogenation process [26]. We have studied the effects of CMHT step on the room-temperature plasticity of TC21 alloy and found that the room-temperature plasticity of TC21 alloy processed by CMHT increased by 93.96% as compared to the original TC21 alloy. However, the effects of hydrogenation temperature on the room-temperature plasticity of CMHT-processed TC21 alloy were not investigated.

Therefore, TC21 alloy was processed by CMHT at different temperatures in the present work. The aim of this work was to investigate the effects of hydrogenation temperature on the room-temperature plasticity of CMHT-processed TC21 alloy. The reasons for improving the room-temperature plasticity were clarified through the characterization of microstructure, which favored the cold forming of TC21 alloy.

2 Experimental

The material used in this study was TC21 alloy bar. The chemical composition and treatment of TC21 alloy specimens were reported in our previous work [27]. The CMHT process was also reported in our previous work [28]. Cylindrical specimens, $d6\text{ mm} \times 9\text{ mm}$, were used for CMHT. The hydrogenation temperature was in the range of 823–1123 K. The initial hydrogen pressure was 10 kPa. Hydrogen of 10 kPa was filled at each interval. The number of filling hydrogen was 4 when the hydrogenation temperature was 823 K, and 7 when the hydrogenation temperature was in the range of 923–1123 K.

Microstructural evolution of TC21 alloy after CMHT was investigated by a YUESHI YM450R optical microscope (OM), a Rigaku D/MAX2500VL/PC X-ray diffractometer (XRD)

and a TECNAI G2 F20 transmission electron microscope (TEM).

Compressive test was carried out on a MTS Landmark universal material experiment machine at room temperature. The compressive speed was 200 mm/min. The plasticity index was determined using different formulas based on the type of fracture exhibited by the TC21 alloy. The plasticity index ε_c was determined using Formula (1) when the specimen fractured obviously. When the specimen had no obvious fracture, the plasticity index ε_r was determined using Formula (2). The microhardness was determined by a HUAYIN HVS-1000A digital microhardness tester with a force of 4.9 N for 30 s.

$$\varepsilon_c = (\varepsilon_b - \varepsilon_{0.2}) \times 100\% \quad (1)$$

$$\varepsilon_r = (h_0 - h_1) / h_0 \times 100\% \quad (2)$$

where ε_b is the ultimate compressive strain when the flow stress reaches its maximum, $\varepsilon_{0.2}$ is the 0.2% yield strain, and h_0 and h_1 are the heights of the TC21 alloy before and after compressive tests, respectively.

3 Results and discussion

3.1 Pressure–composition isotherm curves and thermodynamic parameters

The pressure–composition isotherm curves of TC21 alloy hydrogenated at different temperatures are shown in Fig. 1. The equilibrium hydrogen pressure increased with the increase of the atom ratio of hydrogen to metal (H/M) at various hydrogenation temperatures. When the H/M was the same, the equilibrium hydrogen pressure increased gradually with the increase of hydrogenation temperature. The H/M decreased with the increase of hydrogenation temperature when the equilibrium hydrogen pressure was the same. In addition, it can be seen that the pressure–composition isotherm curves could be divided into three regions when the hydrogenation temperature was in the range of 823–1023 K and two regions when the hydrogenation temperature exceeded 1073 K. During the process of CMHT, hydrogen atoms first dissolved in α and β phases. Phase transformation occurred from α_H phase to β_H phase and δ hydride after the saturation of hydrogen atoms in the alloy. With increasing the amounts of β_H phase and δ hydride, hydrogen atoms continued to diffuse into β_H phase and δ hydride.

Pressure-composition isotherm curves had a plateau region when the hydrogenation temperature was in the range of 823–1023 K. The plateau region was sloped because of the presence of original β phase in the as-received TC21 alloy, and decreased with the increase of hydrogenation temperature. The plateau region disappeared when the hydrogenation temperature exceeded 1073 K, indicating that the hydrogenation temperature reached the critical temperature, and the transformation from α_H phase to β_H phase finished quickly and entered directly into the third region [29]. With the increase of hydrogenation temperature, the mutual solubility of the three phases (α_H phase, β_H phase and δ hydride) in the plateau region increased and the miscibility gap decreased, so the plateau region decreased or even disappeared [30].

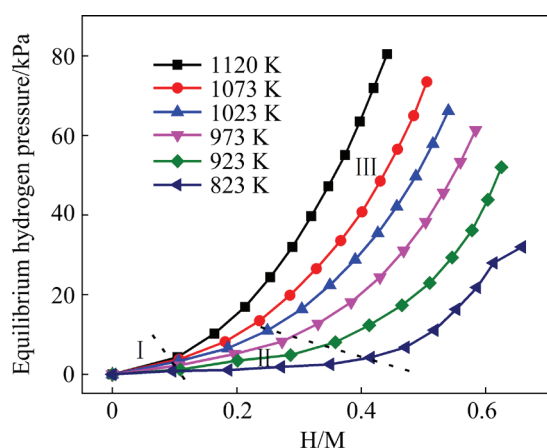


Fig. 1 Pressure-composition isotherm curves of TC21 alloy hydrogenated at different temperatures

The $\ln P-T^{-1}$ curve can be plotted by the Van't Hoff law [31] (as shown in Formula (3)), as shown in Fig. 2. The hydrogenation temperatures and equilibrium hydrogen pressures selected to determine the entropy and enthalpy of TC21 alloy are listed in Table 1. Due to the experimental discrepancy, the hydrogenation temperature and equilibrium hydrogen pressure were taken from the plateau region, and the equilibrium hydrogen pressure was average, as listed in Table 1. According to the fitting results in Fig. 2 and Formula (3), the entropy and enthalpy of TC21 alloy at the plateau region were calculated as $-131.18 \text{ J/(K}\cdot\text{mol)}$ and -54.68 kJ/mol , respectively.

$$\ln p = \frac{\Delta H}{RT} - \frac{\Delta S}{R} \quad (3)$$

where p is the equilibrium hydrogen pressure, T is the hydrogenation temperature, R is the molar gas constant, ΔH is the enthalpy, and ΔS is the entropy.

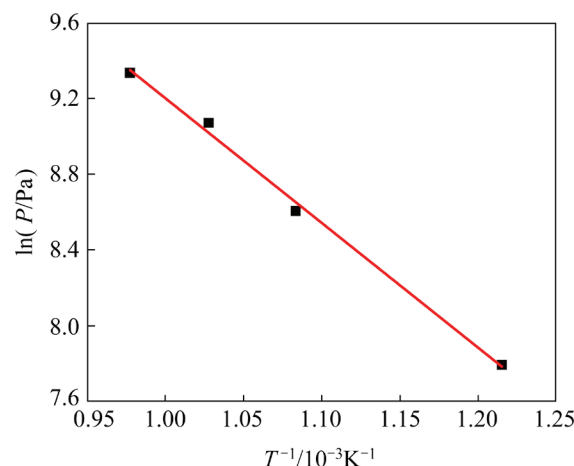


Fig. 2 Curve of $\ln p-T^{-1}$ at plateau region for TC21 alloy

Table 1 Enthalpy and entropy of TC21 alloy at plateau region

Hydrogenation temperature/K	Plateau pressure/Pa	$\Delta H/(\text{kJ}\cdot\text{mol}^{-1})$	$\Delta S/(\text{J}\cdot\text{K}^{-1}\cdot\text{mol}^{-1})$
823	2431.80	-54.68	-131.18
923	5466.22		
973	8697.06		
1023	11314.63		

Sieverts Law can reflect the dissolution ability of hydrogen atoms in TC21 alloy, as shown in Formula 4 [32]. The $H/M-P^{1/2}$ curves can be drawn according to Formula (4), as shown in Fig. 3. It can be seen from Fig. 3 that H/M had a linear relationship with $p^{1/2}$ at each hydrogenation temperature, and increased gradually with the increase of $p^{1/2}$. Sieverts constants of TC21 alloy hydrogenated at different temperatures are shown in Table 2. As can be seen from Table 2, Sieverts constant decreased gradually with the increase of hydrogenation temperature, and decreased by 36.5% when the hydrogenation temperature increased from 823 to 1123 K. Because hydrogen atoms would react with titanium atoms when hydrogen atoms entered into TC21 alloy, and the reaction process was exothermic [33], so the reaction between hydrogen atoms and titanium atoms was inhibited at higher temperatures.

$$K_s = \frac{C}{\sqrt{p}} \quad (4)$$

where K_s is the Sieverts constant, and C is H/M .

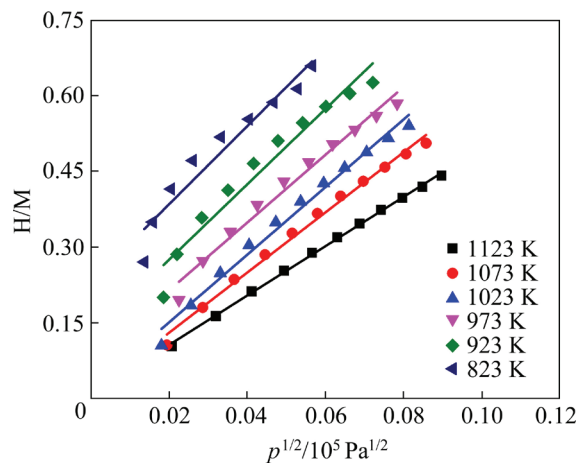


Fig. 3 Curves of $H/M-p^{1/2}$ for TC21 alloy

Table 2 Sieverts constant of TC21 alloy hydrogenated at different temperatures

Hydrogenation temperature/K	Sieverts constant
823	7.72
923	7.45
973	6.70
1023	6.69
1073	5.95
1123	4.90

3.2 Microstructural evolution

3.2.1 OM observation

OM micrographs of the as-received TC21 alloy and alloys treated by CMHT at different temperatures are shown in Fig. 4. Figure 4(a) shows that the as-received TC21 alloy had a typical two-phase structure of α phase (bright) and β phase (dark), and α phase was the main phase. As can be seen from Fig. 4(b), the size of α phase did not change significantly when the hydrogenation temperature was 823 K. As can be seen from the above analysis of the pressure-composition isotherm curves, TC21 alloy was in a state where α_H phase coexisted with β_H phase and δ hydride when the hydrogenation temperature was 823 K, and α_H phase did not transition largely to β_H phase. The optical contrast of α and β phases changed when the hydrogenation temperature exceeded 923 K, as shown in Figs. 4(f, g). The internal stress and crystal defect caused by the precipitation of hydride reduced the electrochemical potential of the α phase [27,34]. However, hydrogen, as a β stable element, could enhance the chemical properties of β

phase and increase the electrochemical potential of β phase [27,34]. In addition, it can be seen that α phase was still the main phase when the hydrogenation temperature was lower than 1023 K. The size of α phase decreased obviously when the hydrogenation temperature exceeded 1073 K. The contrast between α and β phases was completely reversed, that is, α phase became dark and β phase became bright, as shown in Fig. 4(f). β phase became the main phase, because the hydrogenation temperature exceeded the transition temperature of β phase, phase transition of $\alpha \rightarrow \beta$ occurred, leading to the increase of the amount of β phase. As can be seen from Fig. 4(g), the microstructure of TC21 alloy was mainly composed of β phase when the hydrogenation temperature reached 1123 K, and β phase had obvious grain boundaries.

3.2.2 XRD analysis

XRD patterns of the as-received TC21 alloy and alloys treated by CMHT at different temperatures are shown in Fig. 5. As can be seen from Fig. 5(a), the as-received TC21 alloy had diffraction peaks of hexagonal close packed (HCP) α phase and body-centered cubic (BCC) β phase, and the diffraction peak intensity of α phase was much higher than that of β phase, indicating that the amount of α phase was much greater than that of β phase [35], which was consistent with the metallographic microstructure analysis. Compared with the as-received TC21 alloy, the XRD patterns of TC21 alloy changed greatly after CMHT. As can be seen in Figs. 5(b, c), the diffraction peaks of HCP α_2 phase appeared when the hydrogenation temperatures were 823 and 923 K. The presence of α_2 phase indicated that the addition of hydrogen caused the redistribution of alloying elements. The lattice parameters of α_2 phase were $a=b=0.582$ nm and $c=0.459$ nm. The diffraction peak of α_2 disappeared when the hydrogenation temperature was higher than 973 K, as shown in Figs. 5(d–g). The XRD patterns of Figs. 5(d, e) were similar, indicating that there was little difference in phase composition between them. The diffraction peaks of face-centered-cubic (FCC) δ hydride appeared when TC21 alloy was treated by CMHT, and the lattice parameter of δ hydride was 0.440 nm. The diffraction peak intensity of α phase decreased, while the diffraction peak intensity of β phase increased, indicating that β phase replaced α phase as the main phase when the hydrogenation

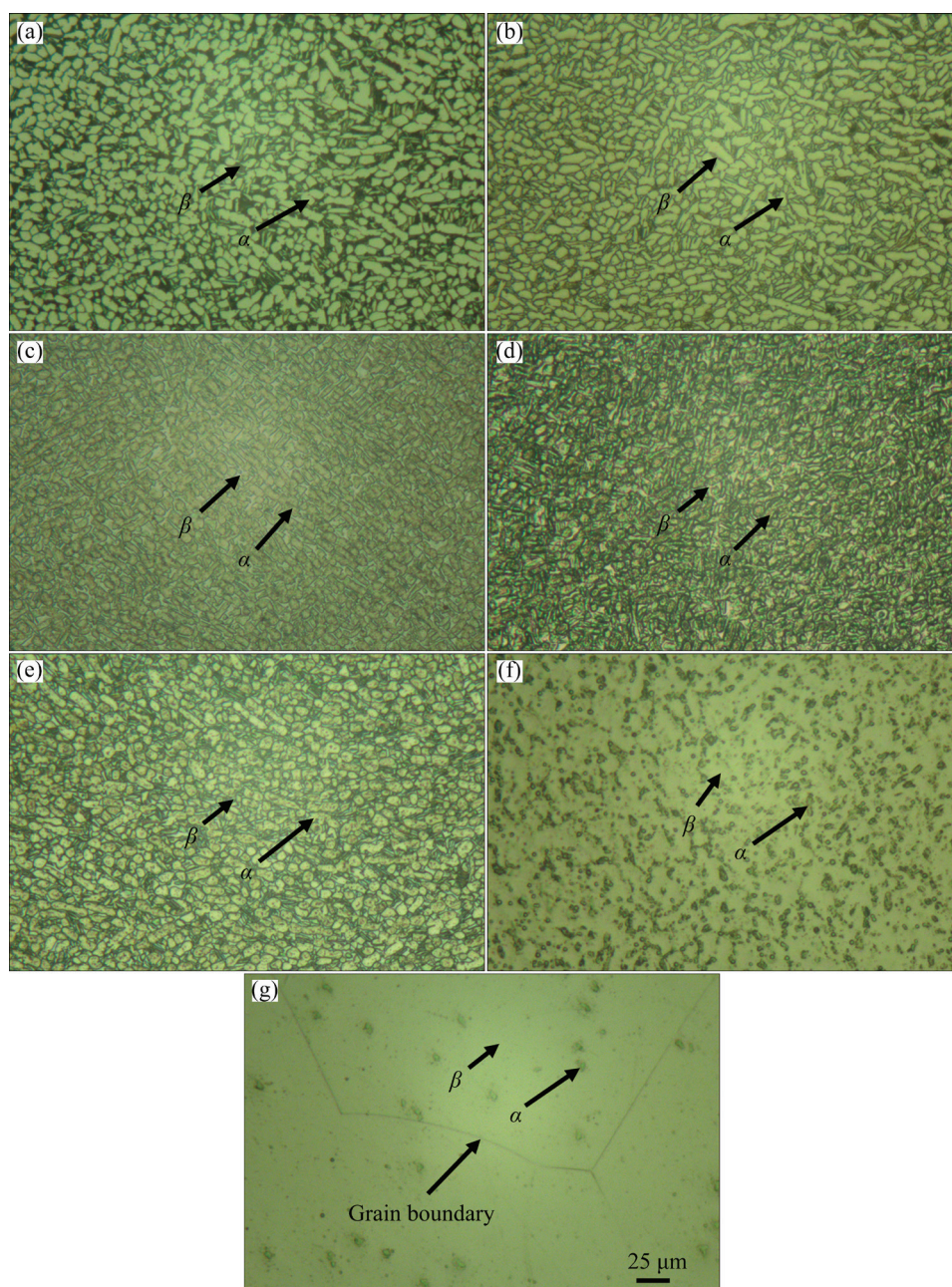


Fig. 4 OM micrographs of as-received TC21 alloy and alloys processed by CMHT at different temperatures: (a) As-received; (b) 823 K; (c) 923 K; (d) 973 K; (e) 1023 K; (f) 1073 K; (g) 1123 K

temperature reached 1073 K. Moreover, the diffraction peak intensity of β phase was much higher than that of α phase when the hydrogenation temperature reached 1123 K.

3.2.3 TEM investigation

The TEM images of the as-received TC21 alloy are shown in Fig. 6. It can be seen from Fig. 6(a) that the as-received TC21 alloy contained α and β phases [27], and the corresponding selected-area electron diffraction (SAED) patterns of α and β phases were $[1\bar{1}0\bar{1}]_{\alpha}$ and $[001]_{\beta}$ zone axes, respectively. Figure 6(b) showed the high

resolution transmission electron microscopy (HRTEM) image of α phase and the corresponding fast Fourier transformation (FFT) diffraction pattern of the selection area which could be further used to observe and confirm the morphology and characteristics of α phase. The $(10\bar{1}1)$ and $(0\bar{1}11)$ lattice planes of α phase are shown in Fig. 6(b). The corresponding inverse fast Fourier transformation (IFFT) in the selection area of Fig. 6(b) is shown in Fig. 6(c). It can be seen from Fig. 6(c) that the interplanar spacing of $(0\bar{1}11)_{\alpha}$ was determined to be 0.219 nm.

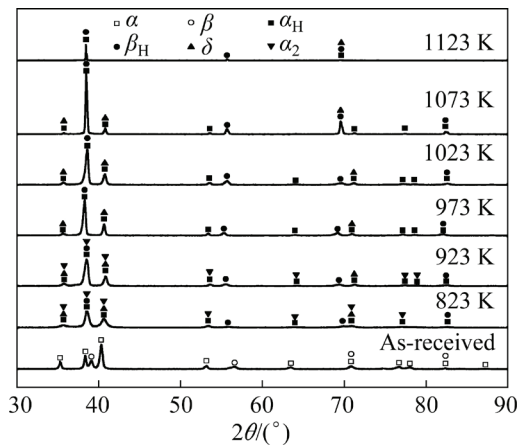


Fig. 5 XRD patterns of as-received TC21 alloy and alloys processed by CMHT at different temperatures

The TEM images of TC21 alloy processed by CMHT at 823 K are shown in Fig. 7. It can be seen from Fig. 7(a) that α_2 phase precipitated in α phase when the hydrogenation temperature was 823 K,

and the corresponding SAED patterns of α phase with $[1\bar{2}1\bar{3}]_{\alpha}$ zone axis and α_2 phase with $[\bar{1}2\bar{1}0]_{\alpha_2}$ zone axis are shown in Fig. 7(a). Figure 7(b) showed that δ hydride formed in α phase, and the crystal band axis of δ hydride was $[\bar{1}10]_{\delta}$. Therefore, it can be concluded that the microstructure of TC21 alloy underwent significant changes compared with the as-received TC21 alloy when the hydrogenation temperature was 823 K.

The TEM images of TC21 alloy processed by CMHT at 923 K are shown in Fig. 8. Figure 8(a) showed that α_2 phase still precipitated in α phase when the hydrogenation temperature was 923 K, which was consistent with the XRD analysis, and the corresponding SAED patterns of α phase with $[000\bar{1}]_{\alpha}$ zone axis and α_2 phase with $[000\bar{1}]_{\alpha_2}$ zone axis are shown in Fig. 8(a). As can be seen from Fig. 8(b), δ hydride still existed in α phase, and the corresponding SAED patterns of δ hydride with $[211]_{\delta}$ zone axis and α phase with $[2\bar{4}2\bar{3}]_{\alpha}$ zone axis

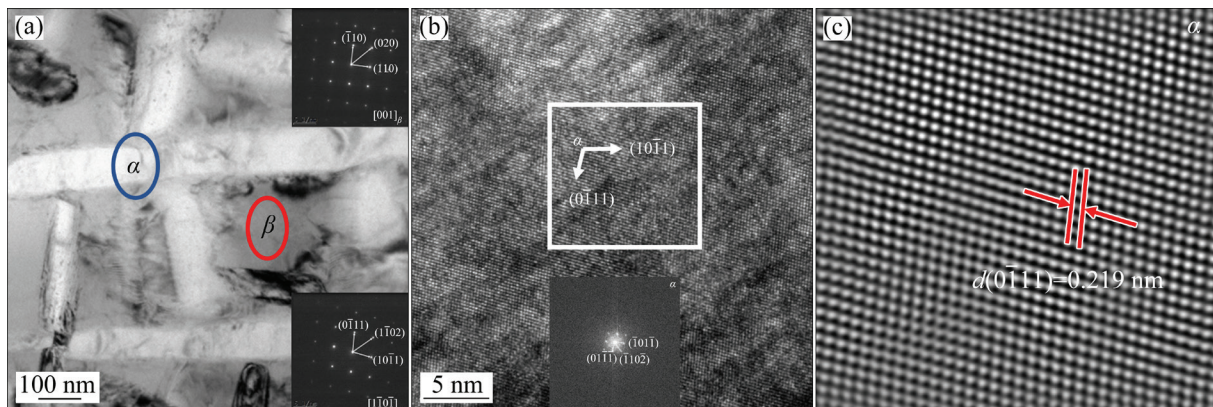


Fig. 6 TEM images of as-received TC21 alloy: (a) α phase and β phase, and corresponding SAED patterns; (b) HRTEM image of α phase and corresponding FFT diffraction pattern; (c) Corresponding IFFT image of α phase

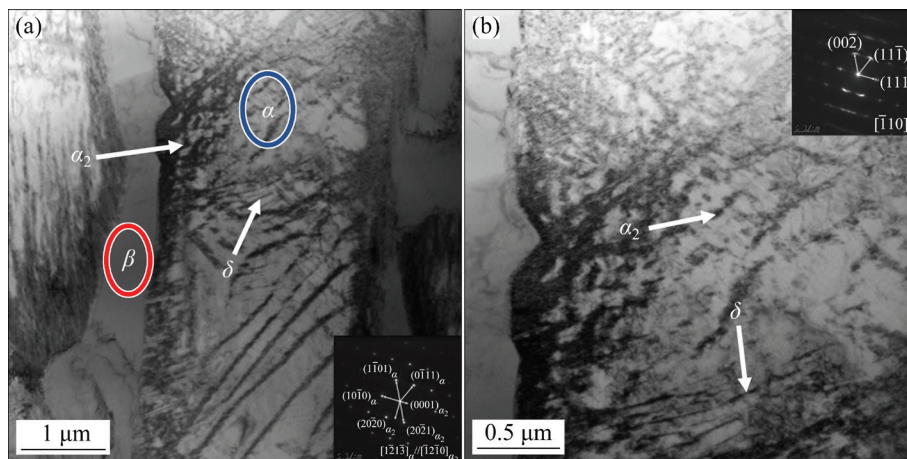


Fig. 7 TEM images of TC21 alloy processed by CMHT at 823 K: (a) α_2 phase formed in α phase, and corresponding SAED pattern; (b) δ hydride formed in α phase, and corresponding SAED pattern of δ hydride

are shown in Fig. 8(b). Figure 8(c) also showed that δ hydride formed in α phase. As can be seen from Figs. 7 and 8, α_2 phase and δ hydride precipitated in α phase when the hydrogenation temperature was lower than 923 K, which could reduce the plasticity of TC21 alloy, as confirmed by the subsequent compression experiments.

The TEM images of TC21 alloy processed by CMHT at 973 K are shown in Fig. 9. Figure 9

showed that a large number of acicular δ hydride precipitated in TC21 alloy when the hydrogenation temperature was 973 K, indicating that this temperature favored the formation of δ hydride. Figure 9(a) showed that a large amount of δ hydride formed in α phase. However, there were still a few δ hydride present in β phase. According to the metallography analysis, α phase remained the main phase when the hydrogenation temperature was lower

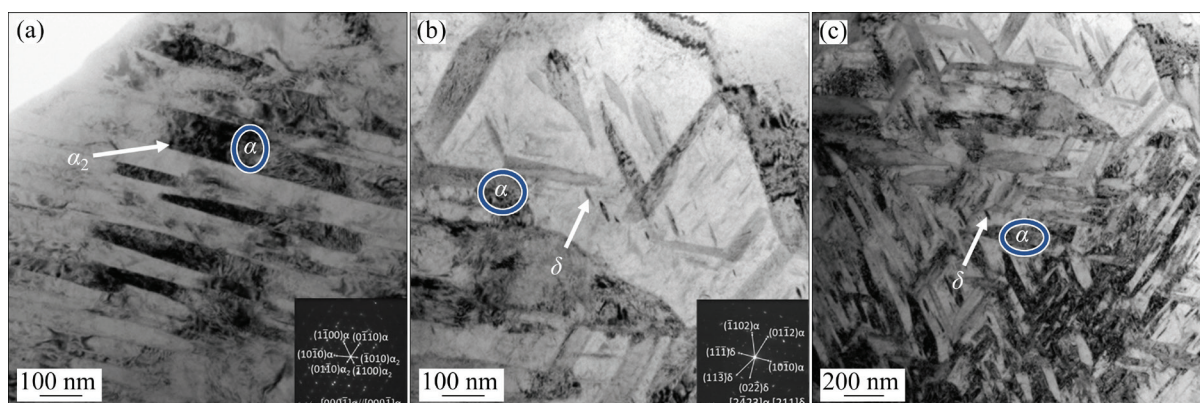


Fig. 8 TEM images of TC21 alloy processed by CMHT at 923 K: (a) α_2 phase formed in α phase, and corresponding SAED pattern; (b) δ hydride formed in α phase, and corresponding SAED patterns; (c) δ hydride and α phase

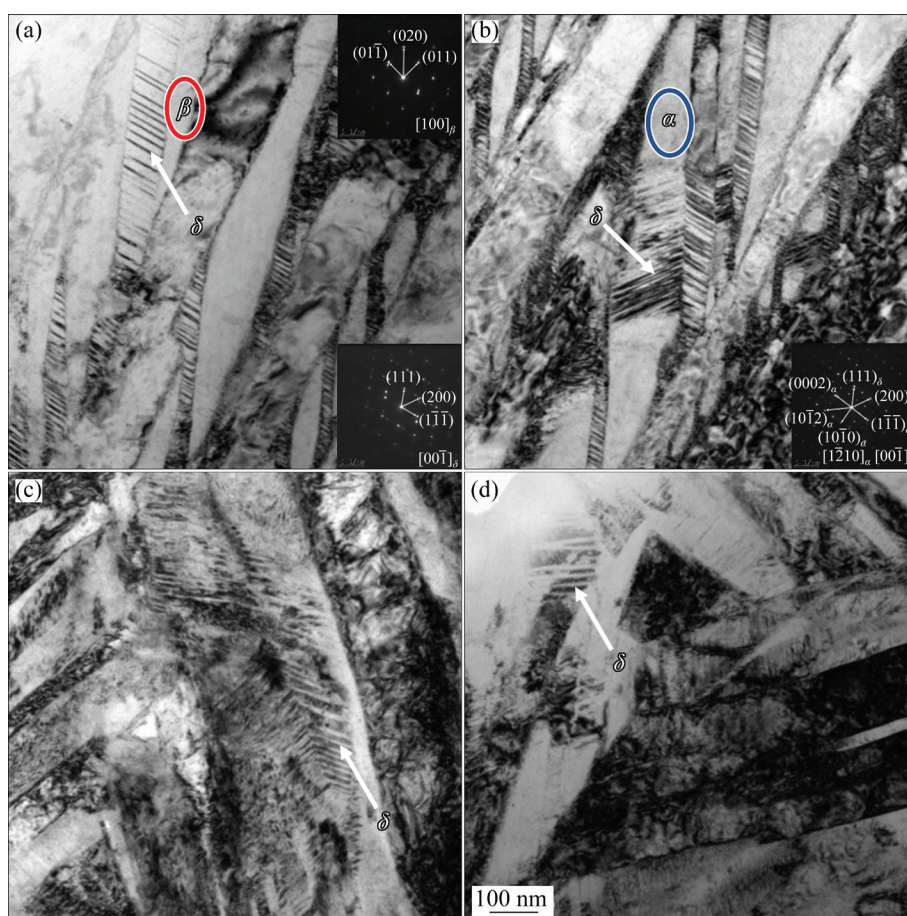


Fig. 9 TEM images of TC21 alloy processed by CMHT at 973 K: (a) δ hydride and β phase, and corresponding SAED patterns; (b) δ hydride formed in α phase, and corresponding SAED patterns; (c, d) δ hydride

than 1023 K, and hydrogen mainly entered into α phase, leading to the precipitation of hydride. In addition, the corresponding SAED patterns of δ hydride with $[00\bar{1}]_\delta$ zone axis and β phase with $[100]_\beta$ zone axis are shown in Fig. 9(a). Figure 9(b) showed the formation of δ hydride in α phase, and the corresponding SAED patterns of δ hydride with $[00\bar{1}]_\delta$ zone axis and α phase with $[1\bar{2}10]_\alpha$ zone axis. Figures 9(c) and (d) also showed many δ hydride existed in TC21 alloy processed by CMHT at 973 K.

The TEM images of TC21 alloy processed by CMHT at 1023 K are shown in Fig. 10. Figure 10(a) showed that there was α phase, β phase and δ hydride in TC21 alloy processed by CMHT at 1023 K. The corresponding SAED patterns of α phase with $[1\bar{2}1\bar{3}]_\alpha$ zone axis and β phase with $[00\bar{1}]_\beta$ zone axis are shown in Fig. 10(a). At 1023 K, α_2 phase disappeared because it had completely transformed into α phase at higher temperature. Figure 10(b) showed δ hydride and the SAED pattern of δ hydride with $[001]_\delta$ zone axis. It can be seen that the amount of δ hydride decreased when

the hydrogenation temperature increased to 1023 K, indicating that δ hydride became unstable at higher temperature. Figure 10(c) showed the image of HRTEM and the corresponding FFT diffraction pattern of zones marked, which could further confirm and observe the morphology and characteristics of δ hydride. The (020) and (200) lattice planes of δ hydride are shown in Fig. 10(c). The corresponding IFFT in the selection area of Fig. 10(c) is shown in Fig. 10(d), and the interplanar spacing of $(020)_\delta$ was determined to be 0.258 nm.

The TEM images of TC21 alloy processed by CMHT at 1073 K are shown in Fig. 11. Figure 11(a) showed that δ hydride formed in α phase, and also revealed the corresponding SAED patterns of δ hydride with $[\bar{1}\bar{1}0]_\delta$ zone axis and α phase with $[\bar{1}2\bar{1}0]_\alpha$ zone axis. Figure 11(b) showed that δ hydride formed from the boundary of β phase and the corresponding SAED pattern of δ hydride was $[\bar{1}12]_\delta$ zone axis. According to Fig. 11(b), δ hydride began to precipitate around the boundary of β phase,

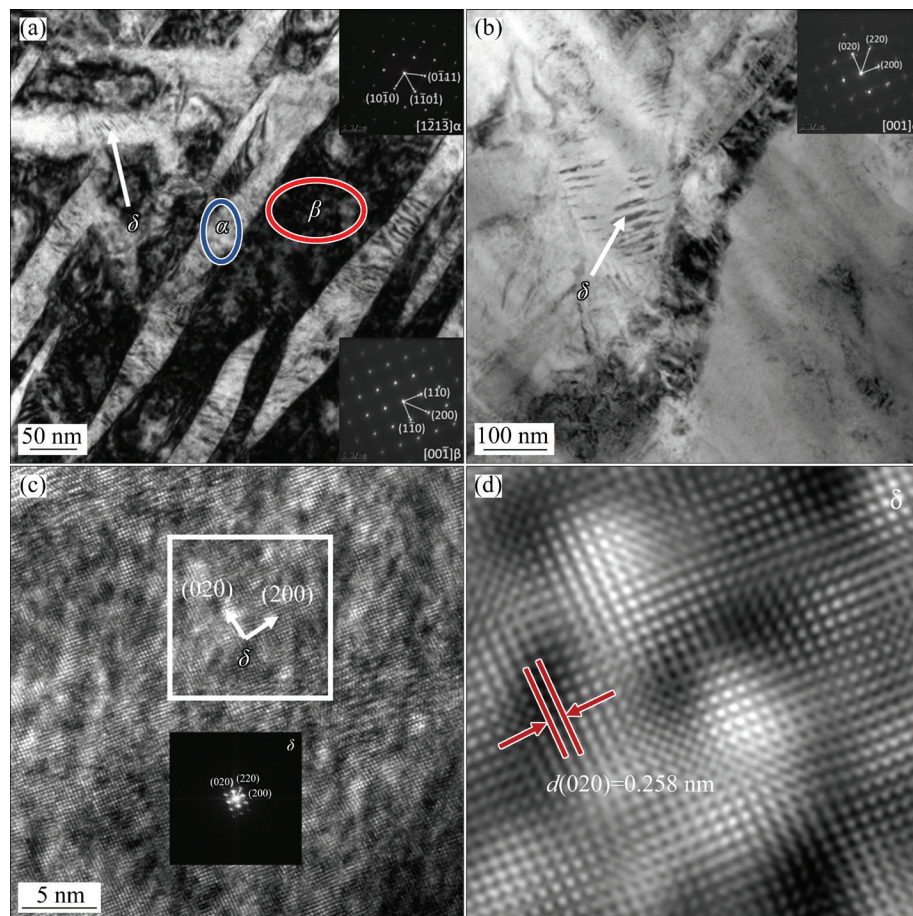


Fig. 10 TEM images of TC21 alloy processed by CMHT at 1023 K: (a) α phase, δ hydride and β phase, and corresponding SAED patterns of α phase and β phase; (b) δ hydride and corresponding SAED pattern; (c) HRTEM image of δ hydride and corresponding FFT diffraction pattern; (d) Corresponding IFFT image of δ hydride

and the amount of δ hydride began to decrease gradually in α phase. With the increase of hydrogenation temperature, the formation of δ hydride required more hydrogen atoms. Therefore, higher hydrogenation temperature was not conducive to the formation of δ hydride.

The TEM images of TC21 alloy processed by CMHT at 1123 K are shown in Fig. 12. Figure 12(a) showed that the amount of α phase decreased greatly in TC21 alloy and β phase became the main phase when the hydrogenation temperature reached 1123 K because of the occurrence of $\alpha \rightarrow \beta$ phase

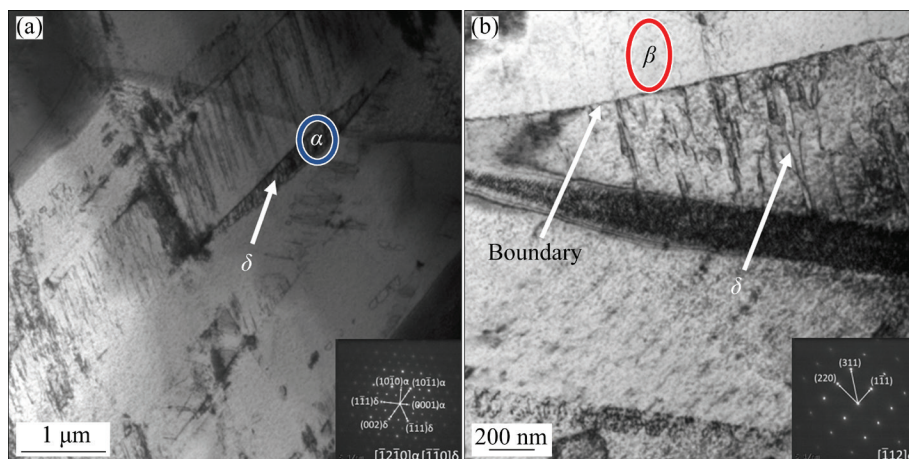


Fig. 11 TEM images of TC21 alloy processed by CMHT at 1073 K: (a) δ hydride formed in α phase, and corresponding SAED patterns; (b) δ hydride formed from boundary of β phase and corresponding SAED pattern

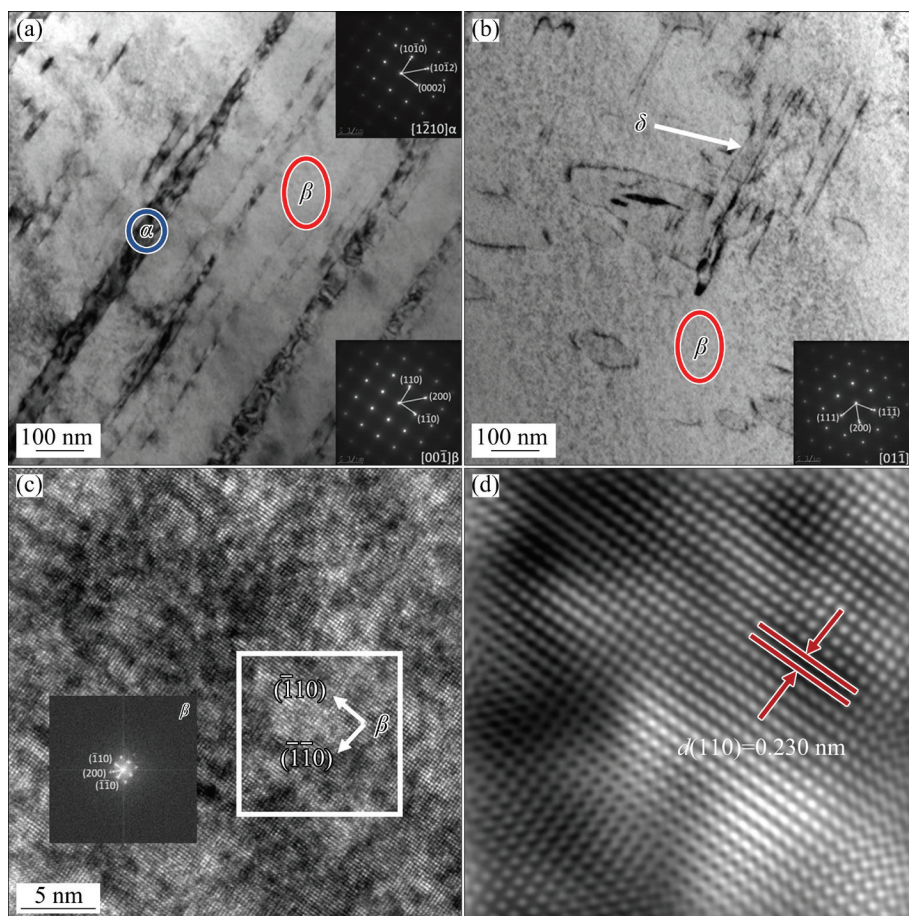


Fig. 12 TEM images of TC21 alloy processed by CMHT at 1123 K: (a) α phase and β phase, and corresponding SAED patterns; (b) δ hydride formed in β phase, and corresponding SAED pattern; (c) HRTEM image of β phase and corresponding FFT diffraction pattern; (d) Corresponding IFFT image of β phase

transformation. Figure 12(a) also revealed the corresponding SAED patterns of α phase with $[1\bar{2}10]_{\alpha}$ zone axis and β phase with $[00\bar{1}]_{\beta}$ zone axis. Figure 12(b) showed that δ hydride formed in β phase when TC21 alloy was hydrogenated at 1123 K. Figure 12(b) also revealed the corresponding SAED pattern of δ hydride with $[01\bar{1}]_{\delta}$ zone axis. It can be seen that the amount of δ hydride increased first and then decreased gradually with the increase of hydrogenation temperature, and reached the maximum at 973 K. δ hydride became unstable with the increase of hydrogenation temperature, which was not conducive to the generation of δ hydride. More information about β phase in Figs. 12(a, b) was obtained by HRTEM, as shown in Fig. 12(c). It can be seen from Fig. 12(c) that the lattice planes of β phase were $(\bar{1}10)$ and $(\bar{1}\bar{1}0)$. The corresponding IFFT in the selection area of Fig. 12(c) is shown in Fig. 12(d). It can be seen from Fig. 12(d) that the interplanar spacing of $(\bar{1}10)_{\beta}$ was determined to be 0.230 nm.

3.3 Mechanical properties

Figure 13 shows the true stress–true strain curves and mechanical properties of the as-received TC21 alloy and alloys processed by CMHT at different temperatures. It can be seen from Fig. 13 that the yield platform was not obvious and the strain hardening was obvious during the compression of the as-received TC21 alloy. The strain hardening phenomenon was obvious when the hydrogenation temperature was lower than 1023 K. The strain softening phenomenon appeared when the hydrogenation temperature exceeded 1073 K and increased with the increase of hydrogenation temperature. The strain softening phenomenon was the most obvious when the hydrogenation temperature reached 1123 K. When the hydrogenation temperature exceeded 1073 K, the number of β phase in TC21 alloy significantly increased, leading to a more noticeable softening phenomenon during the compression process of TC21 alloy.

As shown in Fig. 13(b), it can be seen that the yield strength of TC21 alloy first increased and then decreased, and finally increased again with the increase of hydrogenation temperature. The yield strength of TC21 alloy reached the maximum when the hydrogenation temperature was 823 K, which increased by 30.84% compared with the as-received

TC21 alloy. The yield strength of TC21 alloy reached the minimum when the hydrogenation temperature was 1073 K, which was reduced by 1.35% compared with the as-received TC21 alloy.

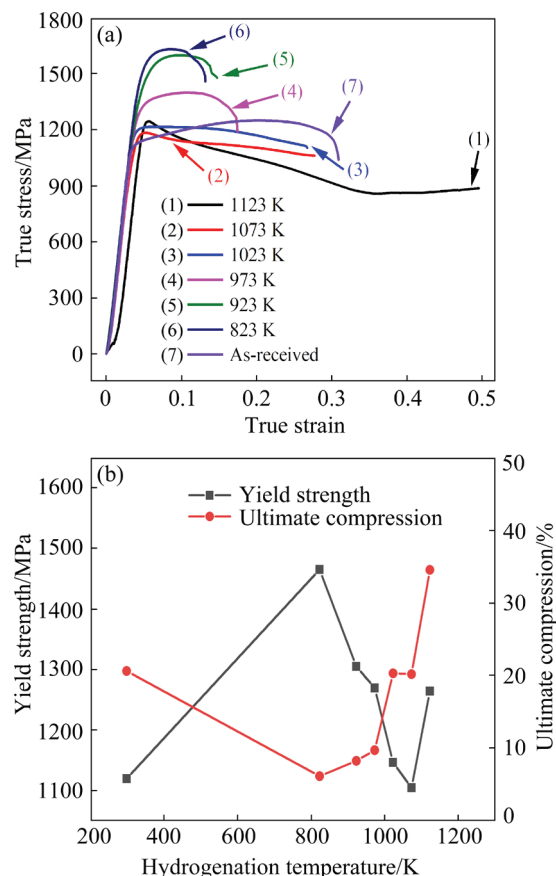


Fig. 13 True stress–true strain curves (a) and mechanical properties (b) of as-received TC21 alloy and alloys processed by CMHT at different temperatures

As shown in Fig. 13(b), the ultimate compression of TC21 alloy decreased first and then increased with the increase of hydrogenation temperature. When the hydrogenation temperature was 823 K, the ultimate compression of TC21 alloy reached the minimum, which was reduced by 70.17% compared with the as-received TC21 alloy. When the hydrogenation temperature was 1123 K, the ultimate compression of TC21 alloy reached the maximum, which increased by 67.51% compared with the as-received TC21 alloy.

It can be seen that the mechanical properties of TC21 alloy processed by CMHT at different temperatures were different obviously, which was related with the change of microstructures in TC21 alloy after CMHT. When TC21 alloy was hydrogenated at 823 and 923 K, a large amount of δ hydride and α_2 phase existed in TC21 alloy, so the

ultimate compression was lower. When the hydrogenation temperature reached 973 K, α_2 phase disappeared, the amount of δ hydride reached the maximum, so the ultimate compression was still lower. When the hydrogenation temperature exceeded 1023 K, the amount of β phase increased and the amount of δ hydride decreased, so the ultimate compression increased. When the hydrogenation temperature reached 1123 K, the amount of β phase reached the maximum, and the amount of δ hydride continued to decrease, so the ultimate compression reached the maximum.

Figure 14 shows the microhardness of the as-received TC21 alloy and alloys processed by CMHT at different temperatures. It can be seen that the microhardness of TC21 alloy increased first and then decreased with the increase of hydrogenation temperature. When the hydrogenation temperature was 823 K, the microhardness of TC21 alloy reached the maximum, which increased by 198.19% compared with the as-received TC21 alloy. When the hydrogenation temperature was 1123 K, the microhardness of TC21 alloy reached the minimum, which decreased by 9.44% compared with the as-received TC21 alloy. When the hydrogenation temperature was 823 K, there were more α_2 phase and δ hydride in TC21 alloy. The microhardnesses of α_2 phase and δ hydride were higher than that of α phase [36], therefore, the microhardness of TC21 alloy hydrogenated at 823 K was larger. The amounts of α_2 phase and δ hydride in TC21 alloy decreased when the hydrogenation temperature increased to 923 K, so the microhardness of TC21 alloy decreased. The microhardness of TC21 alloy continued to decrease when the hydrogenation

temperature exceeded 1023 K, which was caused by the increase of the amount of β phase in TC21 alloy.

4 Conclusions

(1) The equilibrium hydrogen pressure increased with the increase of H/M at various hydrogenation temperatures. Pressure–composition isotherm curves had a plateau region when the hydrogenation temperature was in the range of 823–1023 K. The plateau region disappeared when the hydrogenation temperature exceeded 1073 K.

(2) The entropy and enthalpy of TC21 alloy at the plateau region were calculated as $-131.18 \text{ J/(K}\cdot\text{mol)}$ and -54.68 kJ/mol , respectively. Sieverts constant decreased gradually with the increase of hydrogenation temperature.

(3) The size of α phase decreased obviously and β phase became the main phase when the hydrogenation temperature exceeded 1073 K. The amount of δ hydride increased first and then decreased gradually with the increase of hydrogenation temperature. α_2 phase appeared when the hydrogenation temperatures were 823 and 923 K, and disappeared when the hydrogenation temperature exceeded 923 K.

(4) The ultimate compression of TC21 alloy decreased first and then increased with the increase of hydrogenation temperature. When the hydrogenation temperature was 1123 K, the ultimate compression of TC21 alloy reached the maximum, which increased by 67.51% compared with the as-received TC21 alloy.

CRedit authorship contribution statement

Bao-guo YUAN: Funding acquisition, Resources, Supervision, Writing – Review & editing, Formal analysis; **De-guo QIAN:** Investigation, Data curation, Visualization, Writing – Original draft; **Xiao-xue ZHANG:** Funding acquisition, Resources; **Qiang CHEN:** Conceptualization, Methodology, Supervision; **Ai-chuang TANG:** Investigation, Data curation.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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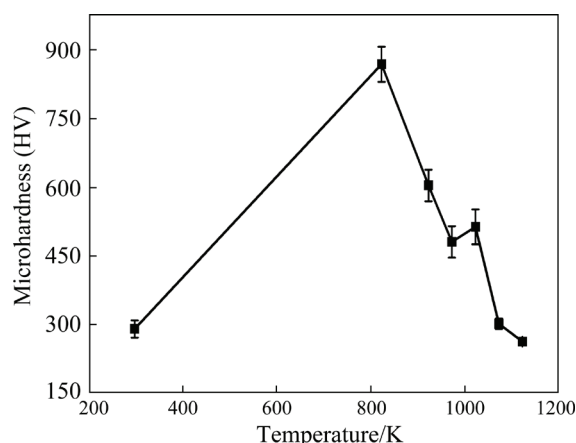


Fig. 14 Microhardness of as-received TC21 alloy and alloys processed by CMHT at different temperatures

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TC21 合金在不同温度下进行连续多步置氢处理时的显微组织及室温力学性能

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摘要: 利用 OM、XRD、TEM 以及压缩试验等方法研究 TC21 钛合金经不同温度连续多步置氢处理后的显微组织及室温力学性能。结果表明: 当氢处理温度超过 1073 K 时, α 相的尺寸明显降低, β 相成为主要相。随着氢处理温度的升高, δ 氢化物的含量呈先增加后逐渐降低的趋势。当氢处理温度超过 1073 K 时, TC21 钛合金在室温压缩变形过程中出现应变软化现象, 且该现象随氢处理温度的升高而越来越显著。随着氢处理温度的升高, TC21 钛合金的极限变形率呈先降低后增加的趋势。当氢处理温度为 1123 K 时, TC21 钛合金的极限变形率比原始合金增加 67.51%。

关键词: TC21 合金; 连续多步置氢处理; 室温力学性能; 压力–组成等温线

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